FATIGUE CRACK TIP DAMAGE-BASED MODELS IN STRUCTURAL PROGNOSIS

S.R. Agnew, Y.J. Ro, M.R. Begley, and R.P. Gangloff
School of Engineering and Applied Science, University of Virginia, Charlottesville, VA, 22904-2442, USA.

ABSTRACT

Accurate structural prognosis demands fundamental models of fatigue crack growth that account for realistic dislocation plasticity, elevated crack tip stresses, and the effect of environment through an environment-affected decohesion law. Conventional plasticity models predict crack tip stress levels too low to explain either substantial H uptake, necessary for H embrittlement, or decohesion of the base metal in vacuum. Non-local plasticity models provide an explanation for highly elevated stress levels at a crack tip. However, there has been little experimental calibration and validation of this modeling. New measurements using electron backscattered diffraction and micro-Laue diffraction with a synchrotron light source aim to characterize the distribution of plasticity very near to the crack-wake surface. Further, characterizing the crystallographic orientation of fatigue crack surface facets assesses the role of H in damage. It is shown that two Al-Cu-based precipitation hardened aluminum alloys exhibit a broad range of facet orientations between {100} and {110} for fatigue in water vapor, with a notable absence of any facets near {111}, which dominated the fracture surface produced in an inert environment.

1 INTRODUCTION

The aluminum alloy airframe provides excellent an example of the challenges facing structural prognosis. The dominant damage modes are time-cycle dependent, including localized corrosion (pitting, intergranular attack and exfoliation) leading to fatigue crack formation from geometrically and chemically altered surfaces, extending to environment sensitive fatigue crack propagation, and ultimately structure failure from instability or mixed mode fracture. For both deterministic and probabilistic fatigue life prediction, crack propagation is described effectively by the relationship between average crack growth rate (da/dN) and the applied or effective stress intensity factor range ($\Delta K = K_{MAX} - K_{MIN}$ or K_{OPEN}). For aerospace aluminum alloys, fatigue crack growth rate depends on a variety of metallurgical, loading and environmental variables [1]. The effect of environment dominates da/dN, as illustrated in Figure 1 [2]. Crack growth rates may be increased by several orders of magnitude by current cyclic loading and exposure to a moist environment relative to ultrahigh vacuum, and this growth rate enhancement is accompanied by one or more transitions in crack path, which signify multiple changes in crack tip damage mechanism. As the environment composition changes from water vapor of varying partial pressure through electrolytes with chloride and other ionic species, da/dN increases with the absolute rate dependent on time through loading frequency [1,3]. Thus, formulation of a fundamental model of da/dN requires focus on the role of environment coupled with cyclic plastic deformation and stress highly localized at the crack tip.

Current da/dN models do not sufficiently capture all aspects of this complex damage scenario. The prevailing three types of model follow. (i) A common form of fatigue crack growth rate model couples rudimentary blunt-crack continuum fracture mechanics descriptions of crack tip plastic strain range or hysteretic energy as the driving force with fatigue life from a Coffin-Manson type formulation from uniaxial fatigue tests [4]. Specific equations relate da/dN to stress intensity range, low-cycle fatigue material life parameters, coefficients from the relationship between crack tip plastic strain and ΔK, and a critical distance. This latter term is often set

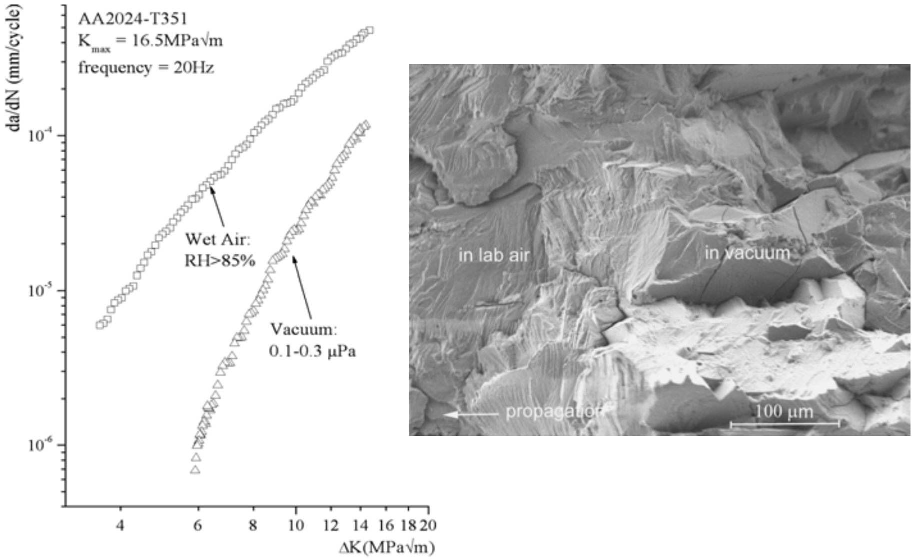

Figure 1: The dominant effect of moist environment on the rate and microscopic damage mechanism of fatigue crack propagation in an aerospace aluminum alloy [2].

adjustably, and microstructure based, on the order of 5-50 μm. While all encompassing, this class of models does not explicitly include environmental effects or sophisticated crack tip mechanics. (ii) A second class of models describes the loading frequency dependence by relating da/dN to water vapor or H transport [1]. From the H diffusion perspective, da/dN is proportional to (H diffusivity/f)$^{1/2}$ and an error function term involving H concentrations [3]. Such models contain adjustable parameters and have not included rigorous crack tip mechanics or H-damage physics. (iii) The final class of models focuses on the details of crack tip fields using continuum plasticity numerical simulations (e.g., dislocation dynamics) coupled to a cohesive model that predicts crack advance in terms of a non-monotonic traction-separation relationship [e.g., 5,6].

A comprehensive approach to modeling environmental fatigue crack growth must account for the critical length scales relevant to the microstructure and H damage mechanism(s), which are likely to be as small as of order 0.1 - 1 μm for crack growth in the near-threshold regime and proportionately larger at higher ΔK. These material-microstructure interactions can be modeled using discrete dislocation dynamics simulations that include dislocation sources, annihilation, interaction with one another and pinning [6]. This approach is attractive since dislocation sources are activated according to local deformation 'needs'; thus, geometrically necessary dislocations (GNDs) are accounted for explicitly. An alternative to full simulation of local dislocation structure and conventional plasticity predictions (which involve no material length scales) are non-local plasticity descriptions, which account for the role of large GND densities due to large strain gradients. These descriptions introduce material length-scales that account for increased stresses due to the presence of dislocations without explicitly tracking individual dislocations. To our knowledge, these formulations have not been applied to cyclic loading scenarios; thus, the influence of strain gradients during cyclic crack tip plasticity is not known. A key consideration is the length-scale over which gradients act, and over which stress elevation is predicted.

The current approach towards structural prognosis involves prediction of da/dN for environmental, as well as vacuum-based, fatigue crack propagation, including several crucial elements. The crack tip mechanics must be formulated across several orders of magnitude of length scale, from the nano to meso scales, to include effects of cyclic loading, microstructure and local elasto-plastic anisotropy, and perhaps dissolved H. The mechanics models must provide plastic strain range distributions, to define the cyclic element of damage, as well as tensile stresses that couple with a cohesive zone model for crack extension. Critically, these mechanics models must be guided and ultimately validated by experimental measurements of deformation parameters local to the crack tip. In parallel with this work, the basic physics of crack tip damage must be understood so that the reason for the preferred crack path transition illustrated in Figure 1 is defined and basis established for a realistic failure criterion.

The present paper outlines current experiments with these objectives. Paralleling this focus on crack tip mechanics [1], it is necessary to probe and model crack (electro)chemistry and the H concentration at the crack tip, however, these aspects are beyond the scope of this paper.

2 CRACK TIP MECHANICS VALIDATION

The crystallographic nature of plastic deformation makes it particularly amenable to diffraction-based characterization approaches. Metal plasticity invariably involves some GNDs in addition to the statistical or incidental dislocations, which actually sustain most of the plasticity. Because the presence of GNDs results in changes in lattice orientation, their presence may be detected by techniques, which are sensitive to lattice orientation. Electron backscattered diffraction (EBSD) has been applied to investigations of plasticity, including that associated with cracking [7]. This technique was applied in a preliminary fashion to a crack branch in an Al-Li alloy (Al-2.5 Cu- 1.6 Li- 0.6 Zn - 0.3 Mn- 0.08 Zr wt%) fatigued in vacuum. The measurements indicate that the extent of plasticity is of order 10 μm, which is larger than a simple plastic zone size prediction, however, a more systematic study must be conducted.

Micro-Laue diffraction using focused synchrotron x-rays probes the material in 3 dimensions with a spatial resolution of ~0.5 x 0.5 x 1.0 μm [8]. For aluminum alloys, it is possible to probe the material to depths of 50-100 μm. The current experiment involves depth resolved line scans perpendicular to the cracking direction in order to probe the effect of stress intensity and the cracking environment on near tip/wake plasticity. Figure 3 illustrates how plasticity is manifested in a diffraction pattern. The undeformed material far from the crack produces sharp, intense spots,

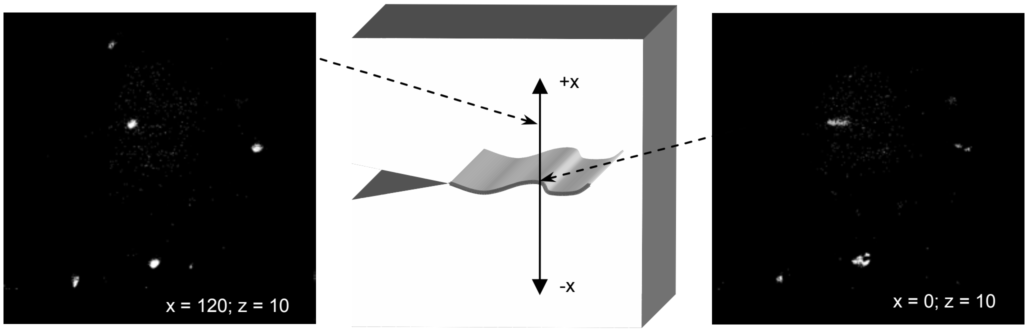

Figure 2: Micro-Laue diffraction patterns obtained from z = 10 μm below the surface. The left pattern having sharp diffraction spots was collected far from the crack (x = 120 μm). The right pattern exhibits asterism of the spots, which is characteristic of plasticity near the crack surface.

while plasticized regions near the crack exhibit substantial asterism. Results from either of these diffraction-based probes have a clear connection with non-local plasticity models grounded in the notion that GNDs associated with gradients in strain are responsible for the increased flow stress in the presence of strain gradients. The GND content is connected with gradients in both orientation and strain [9]. Thus, through correlation of both types of gradient, GND content can be assessed and its impact on the local flow stress predicted for use in and calibration of crack tip mechanics models. An example would be differences in GND content for a crack tip growth in inert vacuum compared to moist nitrogen that produces substantial H dissolved in the zone about the crack tip.

New techniques also enable better assessment of plasticity at the crack tip, thus providing a means to guide and validate non-local plasticity modeling efforts described above. The focused ion beam (FIB) allows preparation of TEM samples for investigation of the dislocation structure at the crack tip/wake. A recent study of a β-Ti alloy demonstrated the capability to prepare TEM foils with electron transparency right through the crack wake up to the crack surface [10]. This allowed assessment of the character of slip in the material (planar vs. wavy) and the typical spacing between dislocation bands. An alternative approach is to *in-situ* strain TEM samples [11], however, such techniques have often been criticized for enforcing unrealistic boundary conditions, uncertain environment and surface film chemistry, and an inability to test within moist environments. Recent advances associated with environmental scanning electron microscopy (ESEM) will allow new *in-situ* cracking studies of more realistic samples and resolution sufficient to probe even nanoscale crack tip geometries [12]. Finally, various deformation mapping techniques [e.g.,13] allow direct determination of strain fields at the crack tip, however, the spatial resolution of strain measurements may not be sufficient to characterize near threshold crack.

3 CRACK PATH CHARACTERIZATION

Determining the dominant mode of decohesion provides a means corroborating models of fatigue crack growth, since the critical inputs to the failure criterion (i.e., normal stress, H content, and intrinsic feature strength) will differ for different failure modes. Transgranular fatigue crack paths may be broadly classified: i) crystallographic; ii) interfacial, iii) striated, per cycle, (iv) damage accumulation followed by advance in *N* cycles. Traditional SEM-fractography (e.g., Fig. 1) has accounted for the majority of investigations of decohesion mechanisms. Additionally, the fine detail of some fatigue fracture surfaces was explored using TEM of replicas. Presently, high resolution SEMs, particularly those with cold-field emission gun sources boast resolution comparable to traditional TEMs and offer advantages: ease of sample preparation, ability to collect data from a statistically significant area of the fracture surface, and greater freedom in sample geometry/thickness, which enables novel *in-situ* experiments..

Substantial evidence establishes that the environmental effect in Figure 1 is explained by hydrogen environment embrittlement and early studies suggested this embrittlement resulted in crystallographic fracture, based upon etch pit or combined stereological and Laue x-ray diffraction analyses [1]. Automated EBSD offers the possibility to determine the crystallographic orientation of a statistically significant number of facets and grains with a resolution previously limited to TEM-based techniques, thus greatly improving upon the qualitative results obtained via etch pit analyses or conventional Laue diffraction [14]. Recent studies demonstrate that environmental fatigue cracking in precipitate strengthened aluminum alloys (Al-4.5Cu-1.5Mg-0.6Mn-0.1Cr and Al-2.5Cu-1.6Li-0.6Zn-0.3Mn-0.08Zr in wt %) appears crystallographic but is not dominated by cleavage-like separation of low-index planes [14,15]. Fatigue cracks are grown in compact tension samples under conditions of constant K_{max} = 16.5 MPa $m^{1/2}$ and descending ΔK [15]. Faceted surfaces were observed for ΔK ranging from threshold to about 7 MPa $m^{1/2}$. The

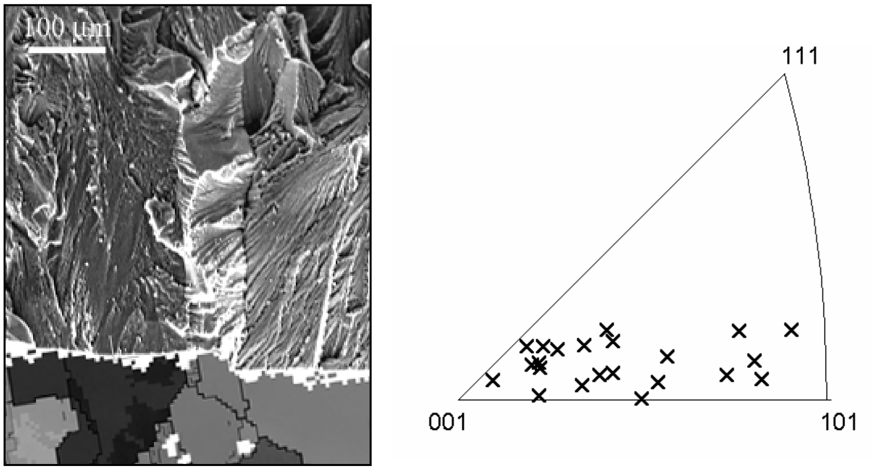

Figure 3: Superposition of an SEM fractograph and OIM microstructure map of an Al-Li-Cu sample fatigue cracked in humid (>85% RH) air shows a clear connection between crack surface morphology and underlying microstructure. The irreducible stereographic triangle at right shows that the orientation of facet normals does not correlate with poles of low index planes.

crystallography of fracture surface facets was investigated using a combined approach involving SEM stereology to determine the spatial orientation of a facet and EBSD of a polished cross-section to determine the crystallographic orientation of the grain in which the facet occurs (Fig. 3). Since both techniques are implemented within a SEM, they may be employed with no intermediate repositioning of the sample, thus reducing uncertainty. The method is capable of facet orientation relative to lattice crystallography with ±2-5° uncertainty depending on the complexity of the crack facet surface [15]. For the example in Figure 3, although the surface has a faceted-crystallographic appearance, there is no tendency for facets preferentially oriented parallel to {001}, {011}, or {111} planes. Rather the facets possess a wide range of orientations spanning the lower half of the irreducible stereographic triangle. The absence of {111} facets is most notable and raises questions regarding the applicability of H enhanced localized plasticity (HELP) type models. This is a particularly meaningful result since the same alloys exhibited crack surface facets parallel or nearly parallel to {111} when stressed cyclically in vacuum. Environment does not exacerbate this tendency for such slip plane related fracture, but rather promotes a new crack morphology. Results of the sort shown in Figure 3 will guide formulation of physically realistic cohesive zone models, including H enhanced decohesion.

4 CONCLUSIONS

First principles, mechanism-based models of environment sensitive fatigue crack growth rate do not exist, despite the fact that they represent a core element of next generation prognosis. Such formulations require a correct description of crack tip stress and plastic strain, including assessment of the importance of strain gradient and discrete dislocation/structure elevation of stress to drive intense H accumulation at a crack tip under cyclic deformation. This description must be validated by high resolution crack tip probes, such as EBSD and 3-D micro-Laue measurements. The basic crack damage mechanism must be defined to enable a meaningful failure criterion, including a cohesive zone formulation. Quantitative definition of prevalent transgranular cracking crystallography by combined stereology and EBSD-based grain orientation is a new capability for this measurement.

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